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ELEVATED-TEMPERATURE FLOW STRENGTH, CREEP RESISTANCE AND DIFFUSION WELDING CHARACTERISTICS OF Ti-6Al-2Nb-1Ta-0.8Mo

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16. Abstract

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SUMMARY

A study of the flow strength, creep resistance and diffusion welding characteristics of the titanium alloy Ti-6Al-2Nb-1Ta-0.8Mo has been conducted. Two mill-processed forms of this alloy were examined. The forged material had been essentially processed above the beta transus (~1275 K) while the rolled form had been subjected to considerable work below the beta transus. Between 1150 and 1250 K, the forged material was stronger and more creep resistant than the rolled alloy. Both forms exhibit superplastic characteristics in this temperature range. Strain measurements during diffusion welding experiments at 1200 K reveal that weld interfaces have no measurable effect on the overall creep deformation. Significant deformation appears to be necessary to produce a quality diffusion weld between superplastic materials. A "soft" interlayer inserted between faying surfaces would seemingly allow manufacture of quality diffusion welds with little overall deformation.

INTRODUCTION

In general, titanium alloys are excellent examples of materials that can be readily diffusion welded (ref. 1). Titanium alloys can have low flow strengths at rather modest temperatures on the order of 1200 K (homologous temperature ≈ 0.63); their high solubility for oxygen and other surface contaminants provides for easy cleaning of faying surfaces at modest temperatures in nonreactive environments; finally, proper thermomechanical processing can induce superplastic behavior in many titanium alloys. Work

reported by Aprigliano and Zanis (ref. 2) has shown that it is possible to diffusion weld the advanced titanium alloy Ti-6Al-2Nb-1Ta-0.8Mo at temperatures ranging from 1090 to 1250 K and low stresses ranging from 0.35 to 3.0 MPa. Low stress levels were of particular interest as they can be produced by atmospheric pressure loading of evacuated vessels. Thus diffusion welding of large diameter titanium pressure vessels was envisioned as an alternative to conventional fusion welding. Aprigliano and Zanis utilized two different mill-processed forms of this alloy, and found they required different diffusion welding schedules (temperature-applied pressure-time) in order to produce satisfactory weldments. They felt that the difference in diffusion welding characteristics of the two mill-processed forms was related to the initial microstructure of the alloys which affected the mechanical flow strength properties under diffusion welding conditions.

The work described herein was undertaken to characterize the mechanical properties in compression of the two mill-processed forms of Ti-6Al-2Nb-1Ta-0.8Mo studied by Aprigliano and Zanis (ref. 2). Compressive flow curves and steady state creep rates were determined over the temperature and stress range of interest for diffusion welding. In addition, several diffusion welds were made, and the quality of the weldments was determined as a function of initial faying surface preparation, deformation across the weld interface, and time under stress and temperature.

EXPERIMENTAL PROCEDURE

Materials and Specimen Geometry

Two mill-processed heats of the titanium alloy Ti-6Al-2Nb-1Ta-0.8Mo were examined in this study: Alpha-beta forged material and Alpha-beta rolled plate material. The forged material was taken from a 0.46 m diameter, 5.1 cm thick, 295 kg ring-rolled cylindrical forging produced by Standard Steel Division of the Baldwin-Lima-Hamilton Corporation. The rolled plate was taken from a 2.5 cm thick plate manufactured by Reactive Metals, Incorporated. Metallography of these materials indicates that the forged alloy had been hot worked above the beta transus (~1275 K) while the rolled alloy had been

subjected to considerable work below the beta transus. The ingots for both mill-processed forms were produced by Reactive Metals, Incorporated. The chemistry of each material is shown on table I. The material used in this study was graciously supplied by L. F. Aprigliano, Naval Ship Research and Development Center.

Right cylindrical compression test specimens were machined out of both mill-processed forms of the alloy. Typical specimen geometries used for flow strength/creep strength determinations are shown in figure 1. Because of the size and quantity of alloy available, 1.4 cm diameter by 4.2 cm length specimens were machined from the forged material, and 1.0 cm diameter by 3.0 cm length specimens machined from the rolled plate. Both flow strength/creep strength specimens possessed a length to diameter ratio of three. Shallow grooves to serve as gage marks were machined in the locations shown in figure 1; in addition, small centering holes (~3 mm dia. by 5 mm length) were machined in the ends of the specimens. The specimen geometry for the diffusion welding work is similar to the 1.0 cm diameter flow strength/creep strength specimen except for its length; the geometry of diffusion welding half specimen is also shown in figure 1. Specimens with gage length parallel to the thickness dimension (5.1 cm) were machined from the forged material, and specimens with gage length parallel to the rolling direction and perpendicular to the rolling direction (long transverse) were machined from the rolled alloy.

Testing and Evaluation

All mechanical property testing and diffusion welding experiments were conducted using a molybdenum compression testing cage. A photograph of the cage with a forged flow strength/creep strength specimen in place is shown in figure 2. Both the top and bottom load transfer connectors were attached to ball-type universal joints to insure uniaxial loading. Thin tungsten metal foil disks were always placed between test specimen and bearing surfaces of the compression cage to prevent diffusion welding of the test specimen to the cage. Testing was conducted in a vacuum of 10^{-5} torr or better in a tantalum resistance heater furnace mounted in a universal testing

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machine. Strains were determined by optically tracking the gage marks through a viewing point with a precision cathometer (precision of each measurement is approximately ±5 μm). During testing, distance measurements were taken starting from the highest gage mark proceeding to the lowest mark; the time of measurement for the highest and lowest gage mark were also recorded. In general, distance measurements for all the gage marks were completed in about 100 seconds. Strains were calculated on the basis of matched pairs of gage marks (highest/lowest, second highest/second lowest, etc.), and the time of the strain measurement was taken as the arithmetic average of the measurement time for the uppermost and lowermost gage marks. Initial gage lengths were determined with a small stress (<0.5 MPa) on the specimen in order to remove all slack in the load train.

Test temperatures were monitored by three Pt/Pt-13% Rh thermocouples which had been spot welded to the specimen near the upper and lower gage marks and the midspan region. Specimens were slowly brought up to temperature over the period of several hours and held, at least, 0.25 hours at temperature before measuring the initial gage lengths and starting the test. In general, the three measured temperatures were quite close and rarely varied by more than ±3 K. Compressive flow strength tests were conducted at constant cross head speed of 8.5×10^{-5} cm/s (strain rate $\sim2.0\times10^{-5}$/s for forged material; strain rate $\sim2.8\times10^{-5}$/s for rolled material). During testing, the applied load was continuously recorded as a function of time. Stress-strain curves were generated by matching the stress-time data to the strain-time data. Compressive stress-strain behavior was determined for both mill-processed forms of alloy at 1150, 1200, and 1250 K. Constant load creep tests and diffusion welding runs were conducted in the load cycle mode on the testing machine with the cross head speed set at 8.5×10^{-5} cm/s. In this mode, the load was cycled between two loading points set at closest possible approach, approximately 1% of full scale setting. Stresses applied in this manner varied only slightly from the average: for example, the actual stress varied from 1.40 to 1.44 MPa during a test designed for a nominal 1.4 MPa stress. Steady state creep rates and final strains were determined directly from strain versus time data. A few creep tests were conducted at 1200 and 1250 K on the

forged material, and at least duplicate creep tests for each test direction were conducted at 1150 and 1200 K on the rolled material. Most creep tests involved a change in load during the test in order to determine steady state creep rates as a function of stress.

Diffusion welding runs were conducted in the exact same manner as creep tests except great care was exercised during initial placement of the half specimens in the compression cage to insure that the faying surfaces were accurately mated. All diffusion welding experiments were conducted at 1200 K and nominally 1.4 MPa. These conditions were selected in order to preserve the original microstructure and to be in a stress range achievable by evacuation of a vessel (ref. 2). Tests were conducted on both lathe finished and ground faying surfaces with total strain being the principal variable. Additionally, several diffusion welding experiments were conducted with a thin (~0.11 cm) interlayer inserted between the half specimens. Following welding a nominally 0.18-cm thick, 1-cm wide, as-welded length slice was electro discharge machined from each as-welded specimen. The slice was then surface ground to nominally 0.15-cm thick and tested in three point bending at room temperature. Bend tests were conducted with the weld interface at the center of bending. The bend fixture had 0.51 cm diameter outside rollers, 0.81 cm diameter center roller, and 1.90 cm span. Maximum bending stresses, S, were calculated from

$$S = \frac{4bP}{wt^2} \tag{1}$$

where b is the half span = 0.95 cm; P is the maximum load; w is the specimen width; and t is the specimen thickness. Bend tests were utilized in this study as previous work (ref. 3) has shown that three point bending with the weld interface at the center of curvature is one of the most severe tests for quality of a diffusion weld.

The microstructures of both mill-processed forms of the Ti-6Al-2Nb-1Ta-0.8Mo alloy were examined before and after creep testing. The weld interfaces of each diffusion welded specimen were also metallographically examined. Specimens for this evaluation were prepared from the welded segments remaining

after EDM removal of the bend test samples. All metallographic specimens were swab etched with a mixture of 5 parts HF, 10 parts HNO_3, and 30 parts lactic acid. Fracture surfaces of the bend specimens were examined in the Scanning Electron Microscope (SEM).

RESULTS AND DISCUSSION

Mechanical Properties

Typical compressive stress-strain diagrams are shown in figure 3 for both forms of the Ti-6Al-2Nb-1Ta-0.8Mo alloy. In addition, a tabular summary of all flow strength data is presented in table II. The stress-strain plots shown in figures 3(a) and (b) are typical of the data generated in this work. In general, the agreement among strains determined from the various gage lengths for each specimen was quite good indicating that the method of evaluating strain was reasonably precise and that deformation was uniform. Maxima in the flow curves were seen for the forged material tested at 1200 and 1250 K and for one 1250 K test of the rolled material. Maxima, strains at which the maximum occurred, and the 0.2 percent offset compressive yield strength data are tabulated in table II. As can be seen in figure 3(c), all the stress-strain diagrams were, at least partially, described by:

$$\sigma = K\epsilon^{n'} \tag{2}$$

where σ is the applied stress; ϵ is the strain; K is the strength coefficient; and n' is the strain-hardening coefficient. Strength coefficients, strain-hardening coefficients, and limits of strain over which equation (2) is followed are given in table II.

From the data in figure 3 and table II, it can be seen that forged material is considerably stronger than the rolled stock over the temperature range examined. While there may be some differences in strength for the rolled material depending on the test direction, the differences are probably not significant.

A typical compressive creep curve involving a change in stress is shown in figure 4 and all tabular creep strength data are presented in table III. The behavior shown in figure 4 is typical of most creep tests in that strain seems to be slightly dependent on the gage length. However, the difference in strain calculated from the two gage lengths tends to be constant over the life of the test, and the steady state creep rates calculated from either gage length were essentially the same. Thus, the slight difference in strain as determined from the two gage lengths was probably due to the difficulty of accurately determining the initial gage lengths and measuring errors during the test rather than non-uniform deformation. For those creep tests involving stress changes, strain rate sensitivity exponents were calculated from

$$\sigma \propto (\dot{\epsilon})^{m} \tag{3}$$

where $\dot{\epsilon}$ is the steady state creep rate and m is the strain rate sensitivity exponent. This exponent is of interest since high m ($\geq$0.3) are indicative of superplastic behavior. The strain rate sensitivity exponents and steady state creep rates as functions of stress and temperature are given in table III. Because of the limited ranges of temperature and stress investigated in this study, no attempt was made to fit the creep data to the normal empirical creep equation.

From the data in table III, it can again be seen that the forged material is much stronger than the rolled alloy. Because of the low creep rates ($\sim 2\times10^{-7}\ s^{-1}$) obtained during testing of the forged material, only a few creep tests were conducted. Diffusion welding of the forged material at low stress levels between 1150 and 1250 K would seemingly be impractical due to the long time required to produce reasonable amounts of deformation ($\sim 5\times10^{4}$ s/percent deformation). The steady state creep rate data for the rolled material indicates that both testing directions have about the same creep strength and that creep rates are fairly reproducible. The strain rate sensitivity exponents appear to indicate that the forged material at 1250 K and the rolled material at 1150 and 1200 K are superplastic.

Typical microstructures of creep tested alloys are shown in figure 5. The microstructures of both alloys were quite uniform. The microstructure in figure 5(a) is identical to that observed in as-received forged alloy; this structure consists of intergranular alpha on the prior beta grain boundaries (quite large equiaxed grains on the order of 2.5 mm dia.), plate like alpha phase which precipitated during hot working, acicular alpha formed by the martensitic transformation of beta, and untransformed beta. The beta phase is the dark constituent while the various alpha phases are light colored. This microstructure indicates that most processing of the forged material was accomplished above the beta transus temperature (~1275 K). When the forged material was exposed to 1250 K, the average size of the alpha constituents increased as can be seen in figure 5(b). The microstructure of the rolled alloy was not affected by exposure to 1150 or 1200 K or creep at these temperatures. The structure visible in figures 5(c) and (d) is identical to that of the as-received alloy and consists of beta and plate-like alpha. The plate-like alpha tended to be somewhat oriented along the rolling direction and each plate was composed of fine grains (~10 μm dia.). No martensitically formed alpha was observed. This microstructure indicates that this material had been subjected to considerable work below the beta transus temperature.

Of the mechanical properties evaluated, the creep strength data can be most readily compared to literature data. As Ti-6Al-2Nb-1Ta-0.8Mo is a mildly beta-stabilized alpha matrix alloy, the most meaningfully strength comparison is with a similar type alloy such as Ti-6Al-4V. Lee and Backofen (ref. 4) have determined the stress versus strain rate properties in tension for small grain size (~6 μm) Ti-6Al-4V between 1073 and 1273 K. The creep rates measured in this study for the rolled material are in general agreement with those determined by Lee and Backofen; although the rolled Ti-6Al-2Nb-1Ta-0.8Mo is somewhat stronger. The forged material is much stronger than Ti-6Al-4V (~ two orders of magnitude difference in creep rate) or the rolled alloy; such differences in strength are probably due to grain size effects as well as type, amount, and distribution of the various phases. The differences in strength observed in the two mill-

processed forms of Ti-6Al-2Nb-1Ta-0.8Mo agrees with the results of Peshkov, Orlova, Ryzokov, and Vorontsov (ref. 5) who observed similar differences in two forms (~10 μm grain size and 400 μm grain size) of the titanium alloy OT4 (Ti-3.2Al-1.6Mn).

Diffusion Welding Studies

Two different diffusion welding geometries were investigated. One couple consisted of compressing mated half specimens machined from the rolled Ti-6Al-2Nb-1Ta-0.8Mo alloy. A typical example of the creep behavior of this geometry is shown in figure 6. The other configuration involved a "soft" interlayer inserted between "hard" half specimens; note that this couple results in two diffusion weld interfaces while the former geometry has only one weld interface. "Soft" interlayers were machined from the rolled material while the "hard" half specimens were machined from the forged alloy. It was expected that only the interlayer would deform during diffusion welding. With one possible exception, essentially no creep deformation was measured during diffusion welding experiments involving interlayers. All creep data from the diffusion welding experiments are presented in table IV. The reported strains and creep rates are based upon the maximum possible gage lengths (~0.75 cm for half specimens; ~2.0 cm for across weld interface).

Comparison of the creep rate data in tables III and IV for the rolled material reveals that the creep behavior under diffusion welding conditions is identical to that under normal creep conditions. The strain data in table IV indicates that the strain across the weld interface is essentially equal to the average of the strains experienced by the half specimens. Thus, it appears that a weld interface does not influence macroscopic creep behavior.

As expected on the basis of creep data in table III, diffusion welding experiments involving the forged material resulted in very little, if any, deformation in each half specimen. Deformation across the weld interface is also quite small and essentially equal to the average strain experienced by

the half specimens. The interlayer and two weld interfaces also apparently do not affect macroscopic strain; this is reasonable as the expected deformation in the interlayer would be quite small even for the longest test (deformation $\approx$ (0.1) cm $\cdot$ 2×10^{-6}/s $\cdot$ 12 500s = 0.0025 cm or about 0.1% overall strain).

The quality of the diffusion welds was evaluated by metallography of the weld interfaces, bend testing of thin slices cut from the as-welded couples, and SEM fractography of failed bend specimens. Typical photomicrographs of weld interfaces are presented in figure 7, and the results of bend testing are given in table V. Plastic deflections at failure were estimated from the plastic (nonlinear) region on load versus cross head motion charts. Typical SEM fractographs of the failed bend specimens are presented in figure 8. The matted areas in figure 8 are regions which were welded and failed by ductile fracture mechanisms; the darker regions in figure 8 are unwelded areas.

Examination of the microstructures in figure 7, bend test data in table V, and fractographs in figure 8 reveals that all the half specimen/half specimen diffusion welds produced in this study are of poor quality. Typically, as much as fifty percent unwelded area can be seen in the fractographs (fig. 8). Such behavior was not expected as previous work (ref. 2) had shown that 0.007 or more strain (as measured by total specimen compression) produces a "good" diffusion weld (as determined by room temperature tensile testing) for this type of geometry. Even the specimen (table IV) which was compressed almost 4% possessed a fair number of weld defects similar to those shown in figures 7(a) and (b) and poor strength.

It appears on the basis of this work that considerably more overall deformation than previously thought (ref. 2) may be required to produce a high quality diffusion weld when diffusion welding superplastic materials. To reduce the overall deformation required for welding, the "hard" half specimen/"soft" interlayer/"hard" half specimen diffusion welding geometry was investigated. As can be seen in table V, the bend properties of welded couples were not equivalent to as-received properties, even after a 12 500 s welding run. On the other hand, the trend is clear that the increasing welding

time at temperature and pressure does improve the quality of the weld both in terms of bend properties, microstructure (fig. 7), and fracture behavior (fig. 8). Thus, this geometry is probably capable of producing good quality diffusion welds with little overall deformation; however, long welding times (>12 500 s) may be required.

With regard to diffusion welding work, the inability to produce sound diffusion welds in the half specimen/half specimen geometry even after about 4 percent deformation was the most disappointing result. This behavior should have, perhaps, been expected on the basis of Hamilton's successful analysis (refs. 6 and 7) of the pressure requirements for diffusion welding of Ti-6Al-4V. For either plain strain or plane stress conditions at the weld interface, his model indicates that the effective stress in the surface asperities, $\bar{\sigma}_A$, is about twice the applied stress, σ. Since steady state creep rates are normally related to stress by

$$\dot{\epsilon} \alpha \sigma^n \tag{4}$$

where n is the stress exponent. The ratio of the strain rates in the asperities to the bulk alloy is

$$\frac{\dot{\bar{\epsilon}}_A}{\dot{\epsilon}} = \left(\frac{\sigma_A}{\sigma}\right)^n \simeq 2^n \tag{5}$$

For normal materials n ranges from 3 to 7, thus equation (5) predicts that most deformation occurring during welding would occur in the asperities. However, for superplastic materials stress exponents are small ($n = 1/m$; based on the data in table III, $n \simeq 1$ for the rolled Ti-6Al-2Nb-1Ta-0.8Mo alloy); consequently the strain rates in the asperities and bulk alloy are not significantly different. Hence, during welding superplastic materials, considerable macroscopic deformation can occur. If the time at stress and temperature is t_b then the sample deformation $\Delta\ell$ after welding is given by

$$\Delta\ell = \left\{2\dot{\bar{\epsilon}}_A H + \dot{\epsilon}(\ell_0 - 2H)\right\} t_b \tag{6}$$

where H is the average asperity height and ℓ_0 is the gage length of the unwelded specimen. Since H is generally quite small (generally less than a few microns) equation (6) reduces to

$$\Delta\ell = \dot{\epsilon}\ell_0 t_b \tag{7}$$

From Hamilton analysis, a sound weld is produced when

$$\dot{\epsilon}_A t_b \simeq \ln 0.5 \tag{8}$$

From equations (5), (7), and (8), the overall strain after diffusion welding is given by

$$\frac{\Delta\ell}{\ell_0} \simeq \frac{\ln 0.5}{2^n} \tag{9}$$

This equation reveals that superplastic materials (n small) will be subject to significant overall deformation during welding while normal materials ($n \geq 3$) will undergo much less deformation.

Experimentally, two studies (refs. 5 and 8) of diffusion welding in Ti alloys have shown that significant overall deformation is required for quality weldments. Peshkov, Orlova, Ryzhkov, and Vorontsov (ref. 5) studied diffusion welding in an approximately 10 μm grain size Ti-3.2Al-1.6Mn alloy. Utilizing room temperature tensile testing as a means of assessing the quality of the diffusion weld, they found that approximately 4 percent deformation at 1173 and 1223 K was required in order to produce welds with tensile strengths equivalent to the as-received alloy. However, even after 9 percent deformation during welding at 1173 K, tensile failure apparently occurred at the weld interface; while 4.5 percent or more deformation during welding at 1233 K resulted in weldments with parent metal strength and ductility. When a more severe test of weld quality involving notch Charpy impact strength testing was used, Peshkov, Orlova, Ryzhkov, and Vorontsov found even more deformation during diffusion welding was

necessary to match parent metal properties. Similarly Shorshorov, et al. (ref. 8) found that large deformations (generally greater than 10%) were required to produce quality diffusion welds as determined by impact testing in several Ti alloys (Ti-3.2Al-1.6Mn, Ti-6Al-4V, Ti-3.5Al-7.5Mo, and Ti-3.5Al-11Mo).

Experimentally and theoretically, it appears that large overall deformations are required to produce quality diffusion welds in alloys which exhibit superplastic behavior. Where large deformations are not possible or desirable, the "soft" interlayer concept offers possibilities as an effective means of obtaining satisfactory diffusion welds. Additionally, it is possible (ref. 1) to diffusion weld in a closed die system which will prevent large deformations except at the weld interface.

In light of the above arguments, we are unable to explain Aprigliano and Zanis's success in diffusion welding Ti-6Al-2Nb-1Ta-0.8 alloys with small overall deformations (ref. 2). It is possible that the use of the more severe test (bend testing) of weld quality found defects that also existed in welds produced by Aprigliano and Zanis; however, examination of their photomicrographs of weld interfaces indicates that their welds are of decidedly better appearance than those shown in figures 7(a) and (b).

SUMMARY OF RESULTS

A study of the elevated temperature mechanical properties and diffusion welding characteristics of two mill-processed forms of Ti-6Al-2Nb-1Ta-0.8Mo has shown the following:

1. The mill-processed form of the alloy involving hot work accomplished above the beta transus is stronger and more creep resistant between 1150 and 1250 K than the mill-processed form involving considerable hot work below the beta transus.

2. Both forms of the alloy exhibit superplastic characteristics.

3. Diffusion weld interfaces do not influence overall creep behavior.

4. Significant amounts of deformation (>4 percent) are seemingly necessary to produce quality diffusion welds between superplastic alloys.

5. Production of quality diffusion welds without external constraints and with little overall deformation appears to be possible through use of a soft interlayer between faying surfaces.

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TABLE I. - NOMINAL COMPOSITION OF FORGED AND ROLLED ALLOYS (REF. 2)

Mill-processed form	Composition, weight percent									
	Al	Nb	Ta	Mo	C	H	N	O	Fe	Ti
Forged	6.0	2.47	1.2	0.79	0.02	0.0055	0.006	0.068	0.07	Bal.
Rolled	6.1	1.9	0.88	0.88	0.02	0.0036	0.009	0.0088	0.05	Bal.

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TABLE II. - COMPRESSIVE STRESS-STRAIN PROPERTIES OF

Ti-6Al-2Nb-1Ta-0.8Mo ALLOYS

Temperature, K	0.2 Percent offset yield stress, MPa	Maximum strength, MPa	Strain at maximum	Strength coefficient, MPa	Strain hardening coefficient	Strain limits	
						Lower	Upper
Forged material (strain rate $\approx 2.0\times10^{-5}$/s)							
1150	26.6	----	-----	171	0.329	0.001	>0.007
1150	27.0	----	-----	203	.367	.001	>.007
1200	16.8	21.2	0.017	65.7	.259	.003	.013
1200	13.6	20.3	.017	71.6	.289	.002	.012
1250	6.4	9.8	.015	114	.55	.0018	.012
1250	7.0	7.4	.005	-----	-----	------	-----
Rolled material - longitudinal testing direction (strain rate $\approx 2.8\times10^{-5}$/s)							
1150	5.5	----	-----	49.5	0.341	0.002	>0.027
1150	6.2	----	-----	55.2	.360	.0024	>.027
1200	3.7	----	-----	17.5	.28	.0025	>.016
1200	3.7	----	-----	13.9	.22	.003	>.018
1200	3.2	----	-----	14.2	.26	.003	>.025
1250	1.2	1.5	0.016	9.6	.416	.003	.016
Rolled material - transverse testing direction (strain rate $\approx 2.8\times10^{-5}$/s)							
1150	8.4	----	-----	39.0	0.287	0.0014	>0.021
1150	7.8	----	-----	39.0	.298	.0016	>.029
1200	2.8	----	-----	20.0	.320	.0013	>.021
1200	2.5	----	-----	20.2	.351	.0016	>.014
1200	4.9	----	-----	14.9	.195	.0028	>.020
1250	1.4	----	-----	8.0	.343	.0034	>.025
1250	1.3	----	-----	5.6	.288	.0019	>.023

TABLE III. - CREEP STRENGTHS OF Ti-6Al-2Nb-1Ta-0.8Mo ALLOYS

Temperature, K	Stress, MPa	Steady state creep rate, s^{-1}	Strain rate sensitivity exponent
Alpha-beta forged material			
1200	5.62	3.0×10^{-7}	1.11
1250	1.39	1.8×10^{-7}	
	2.85	4.0×10^{-7}	
1250	2.85	4.1×10^{-7}	
Alpha-beta rolled material - longitudinal testing direction			
1150	1.54	6.6×10^{-7}	0.73
	2.93	1.6×10^{-6}	
1150	1.54	2.7×10^{-6}	2.71
	2.94	3.4×10^{-6}	
1150	1.54	3.0×10^{-6}	2.04
	2.91	4.1×10^{-6}	
1200	1.47	2.9×10^{-6}	0.84
	2.91	6.2×10^{-6}	
1200	1.47	3.0×10^{-6}	
Alpha-beta rolled material - transverse testing direction			
[a]1140	1.43	5.4×10^{-7}	1.28
	2.85	9.3×10^{-7}	
1150	1.42	1.6×10^{-6}	.98
	2.80	3.2×10^{-6}	
1200	1.38	2.7×10^{-6}	.735
	2.76	6.9×10^{-6}	
1200	1.42	4.1×10^{-6}	1.18
	2.82	7.3×10^{-6}	
1200	1.48	4.3×10^{-6}	

[a]Test originally scheduled to be conducted at 1150 K; however, due to balancing error in temperature recording unit, actual test temperature was 11 K low.

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TABLE IV. - CREEP DATA FROM 1200 K DIFFUSION WELDING EXPERIMENTS

Stress, MPa	Time of test, s	Testing direction	Surface finish faying surfaces	Steady state creep rates, s^{-1}			Compressive strain at end of test		
				Upper half specimen	Lower half specimen	Across weld interface(s)	Upper half specimen	Lower half specimen	Across weld interface(s)
Half specimen/half specimen geometry									
1.45	4 600	Longitudinal	Lathe turned	4.0×10^{-6}	2.7×10^{-6}	3.4×10^{-6}	0.0218	0.0132	0.0183
1.47	5 000	Transverse	Lathe turned	2.0×10^{-6}	3.9×10^{-6}	3.2×10^{-6}	.0107	.0208	.0175
1.49	3 500	Transverse	Lathe turned	3.2×10^{-6}	2.6×10^{-6}	3.1×10^{-6}	.0094	.0096	.010
1.64	15 700	Transverse	Lathe turned	1.7×10^{-6}	2.0×10^{-6}	1.9×10^{-6}	.031	.042	.0374
1.50	6 500	Transverse	Ground	2.8×10^{-6}	2.0×10^{-6}	2.4×10^{-6}	.0197	.0187	.0189
Half specimen/interlayer/half specimen geometry									
1.43	6 500	(a)	Ground	$<1.3\times10^{-7}$	$<1.3\times10^{-7}$	$<1.3\times10^{-7}$	0.0009	0.0009	0.0009
1.42	5 900	Transverse[b]	Ground	$<1.\times10^{-7}$	$<2.6\times10^{-7}$	$<2.1\times10^{-7}$	.0006	.0015	.0012
1.42	2 150	Transverse[b]	Ground	5.5×10^{-7}	5.6×10^{-7}	4.4×10^{-7}	.0029	.001	.0017
1.42	12 500	Transverse[b]	Ground	$<1.1\times10^{-7}$	$<1.3\times10^{-7}$	$<1.3\times10^{-7}$	.0014	.0017	.0016

[a]No interlayer.

[b]Testing direction of interlayer.

TABLE V. - BEND STRENGTH AND DUCTILITY OF

DIFFUSION WELDED SPECIMENS

Prior strain across interface	Maximum load, N	Maximum stress, MPa	Plastic deflection at failure, cm	Comments
Half specimen/half specimen couples				
None-as received rolled material	2025	2867	1.63	Pulled through bend fixture without failing
0.01	725	1062	0	Failed at weld
.0175	1388	2032	0	Failed at weld
.0183	792	1165	0	Failed at weld
.037	1415	2092	0.022	Failed at weld
.0189	975	1450	0	Failed at weld
Half specimen/interlayer/half specimen couples				
None-as received forged material	1709	2407	0.229	
	1945	2619	0.268	
[a,b]6500	303	483	0	Failed at weld
[a]2150	596	944	0	Failed at weld
[a]5900	1028	1773	0.01	Failed at weld
[a]12 500	1326	2226	0.127	Failed through part of interlayer and weld interface

[a]Time of diffusion weld, s.

[b]No interlayer.

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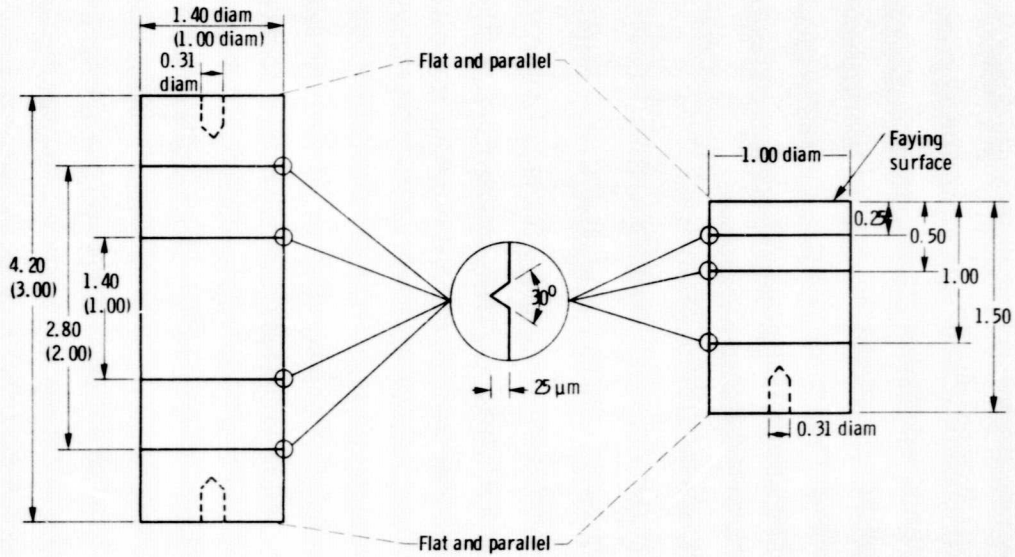

(a) Flow strength/creep strength specimen.

(b) Diffusion welding half specimen.

Figure 1. - Test specimen geometry. All dimensions in centimeters unless noted.

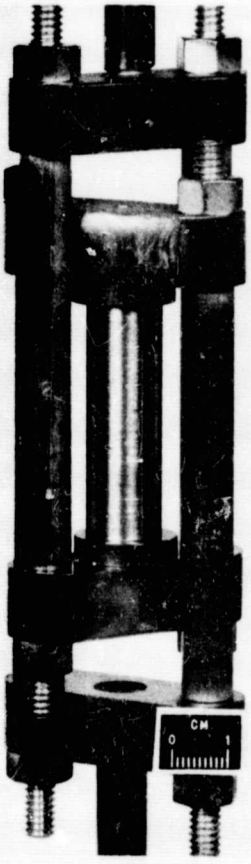

Figure 2. - Compression test cage and flow strength/creep specimen (forged material) in place.

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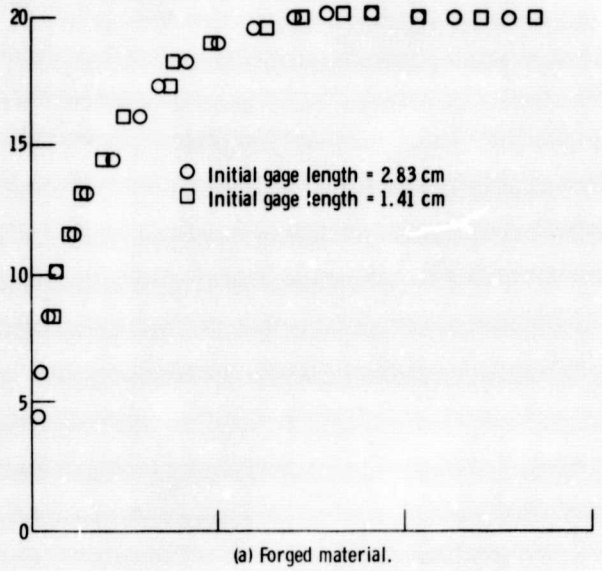
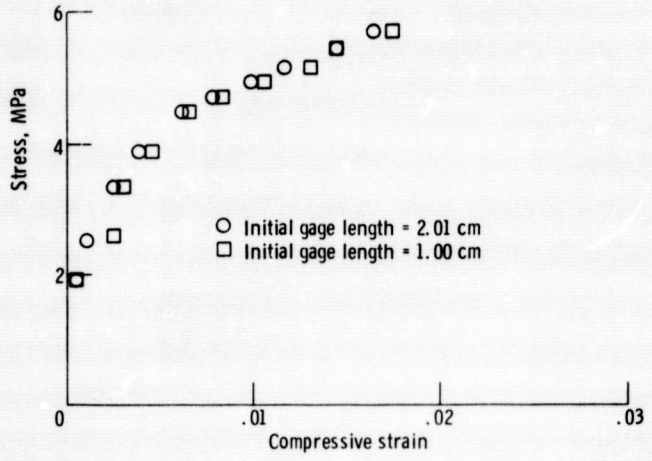
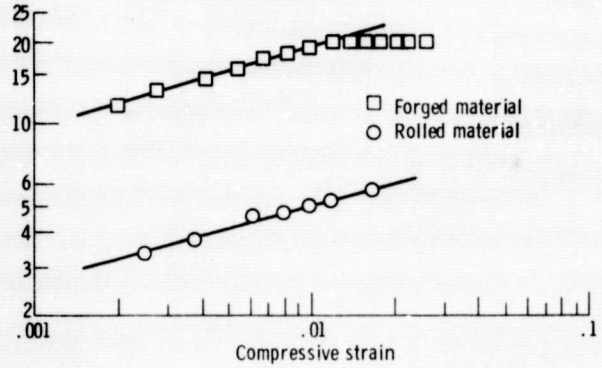


(a) Forged material.



(b) Rolled material - longitudinal test direction.



(c) Logarithmic stress-strain curves for data shown in (a) and (b).

Figure 3. - Typical compressive stress-strain diagrams for Ti-6Al-2Nb-1Ta-0.8Mo alloys tested at 1200 K.

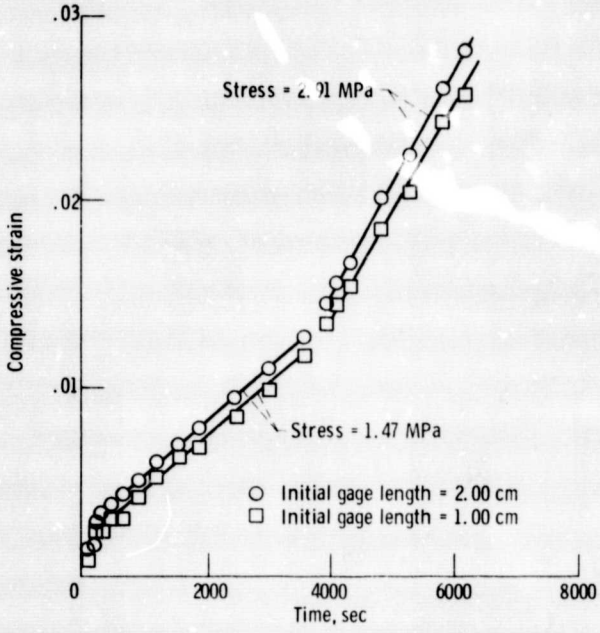
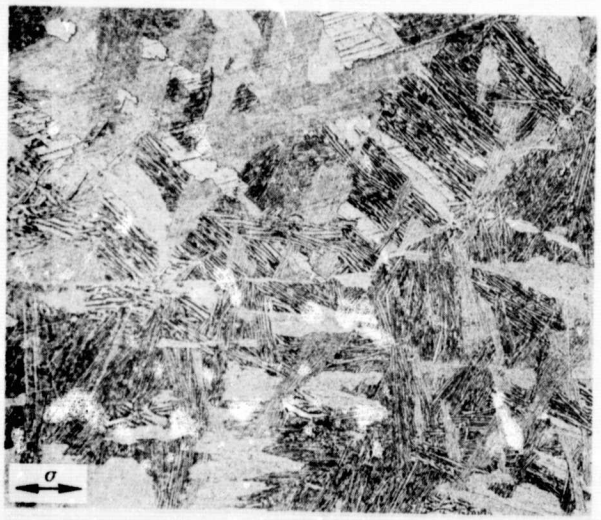
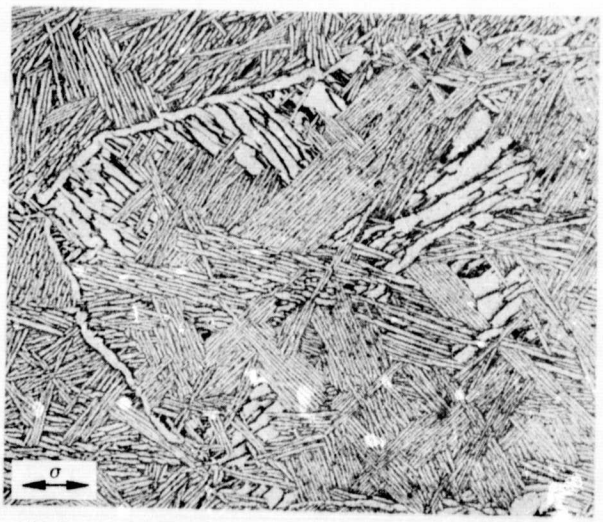


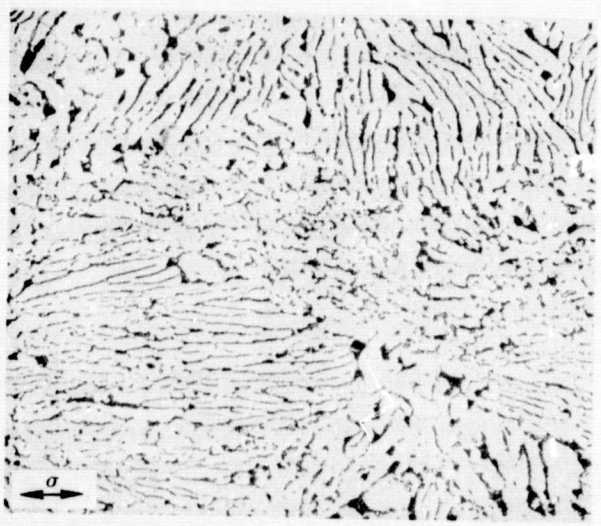
Figure 4. - Typical creep curve involving a change in stress for rolled Ti-6Al-2Nb-1Ta-0.8Mo material tested in longitudinal direction at 1200 K.



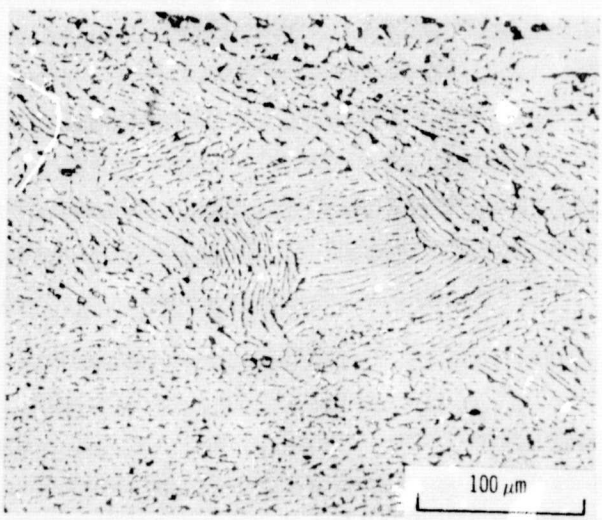
(a) Forged material; creep tested at 1200 K and 5.62MPa for 7300 seconds to 0.0028 strain.



(b) Forged material; creep tested at 1250 K and 2.85MPa for 7350 seconds to 0.0044 strain.



(c) Rolled material; creep tested in longitudial direction at 1200 K - 1.47/2.91MPa for 6200 seconds to 0.028 strain.



(d) Same as part (c). Perpendicular to stress axis.

Figure 5. - Typical photomicrographs of creep tested Ti-6Al-2Nb-1Ta-0.8Mo alloys.

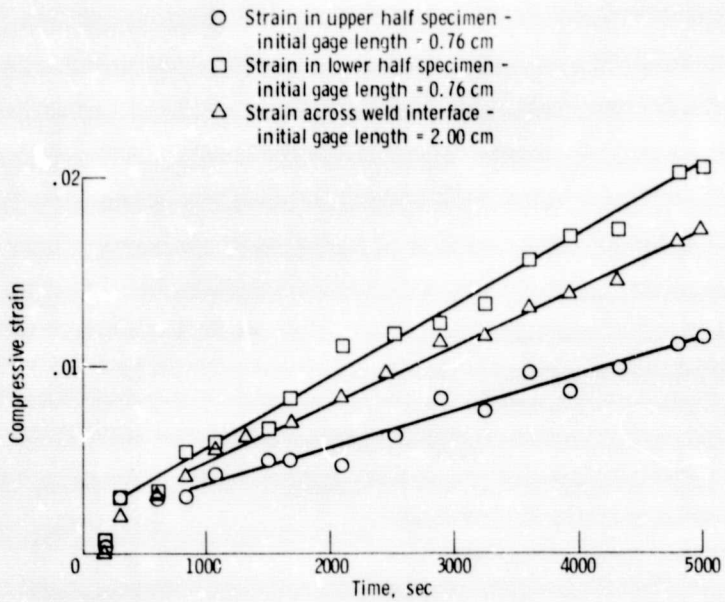
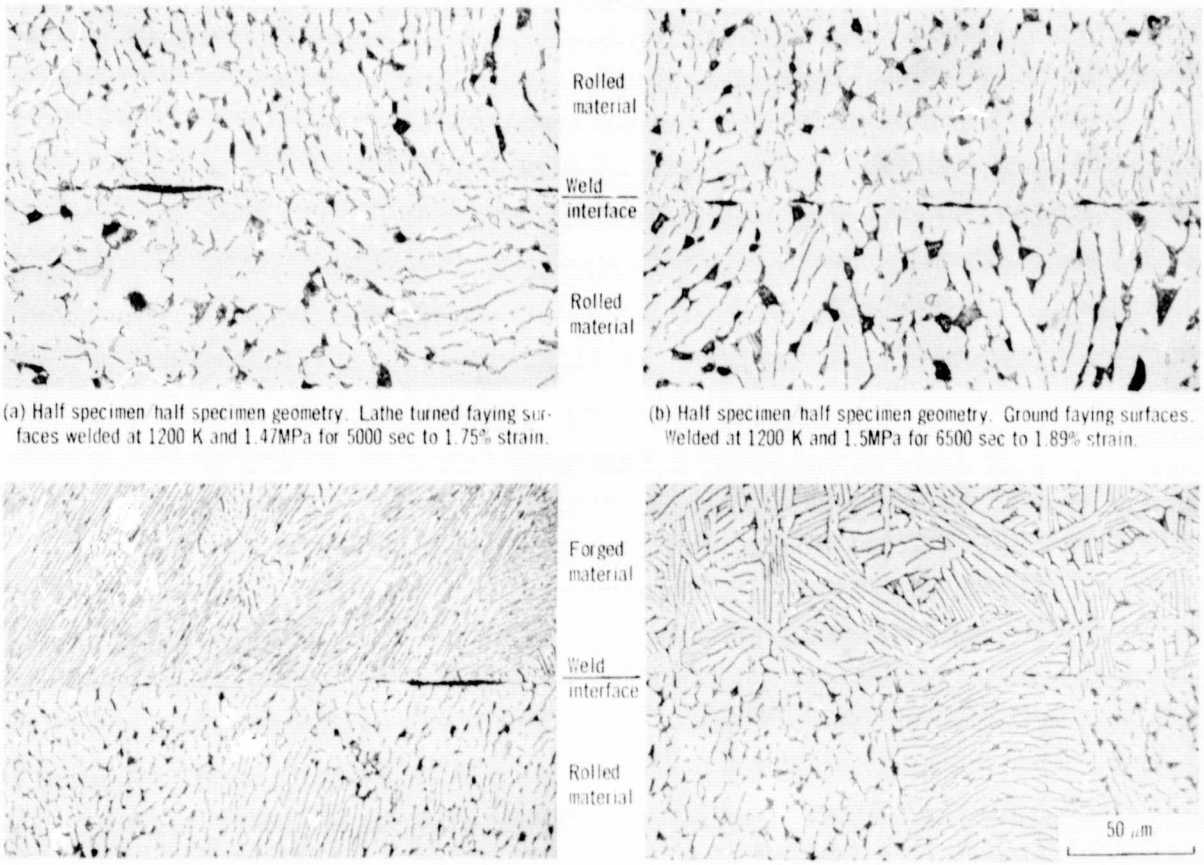


Figure 6. - Typical creep behavior of a mated half-specimen couple during a diffusion welding experiment at 1200 K and 1.47 MPa. Rolled material; transverse direction.



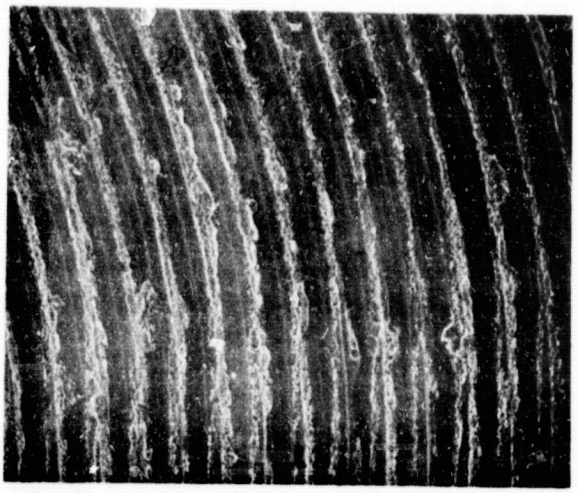
(a) Half specimen–half specimen geometry. Lathe turned faying surfaces welded at 1200 K and 1.47MPa for 5000 sec to 1.75% strain.

(b) Half specimen–half specimen geometry. Ground faying surfaces. Welded at 1200 K and 1.5MPa for 6500 sec to 1.89% strain.

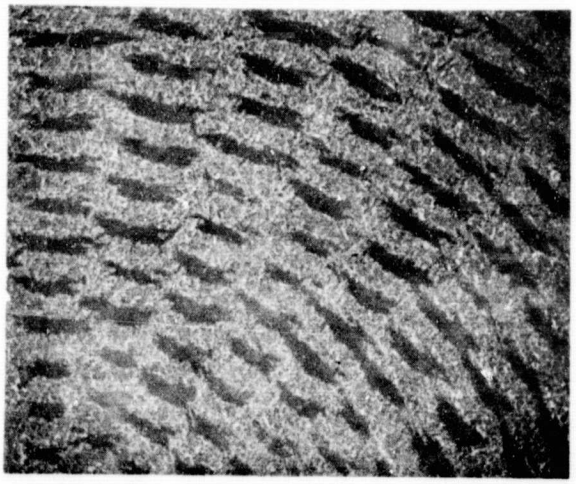
(c) Half specimen–interlayer–half specimen geometry. Ground faying surfaces welded at 1200 K and 1.42MPa for 5900 sec to 0.12% strain.

(d) Half specimen–interlayer–half specimen geometry. Ground faying surfaces welded at 1200 K and 1.42MPa for 12 500 sec to 0.16% strain.

Figure 7. - Typical photomicrographs of the weld interfaces. Applied compressive stress in vertical direction.



(a) Half specimen/half specimen geometry. Diffusion welded at 1200 K and 1.49MPa for 3500 sec to 0.01 strain. Lathe turned faying surfaces.



(b) Half specimen/half specimen geometry. Diffusion welded at 1200 K and 1.64MPa for 12 500 sec to 0.037 strain. Lathe turned faying surfaces.

Figure 8. - Typical SEM micrographs of the fracture surfaces of failed bend test specimens machined from diffusion weld couples.

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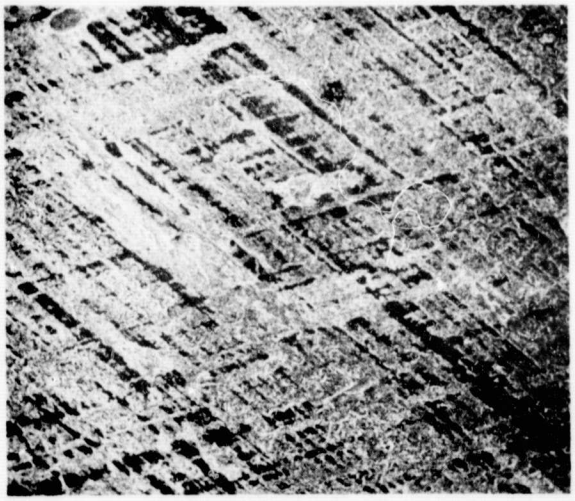
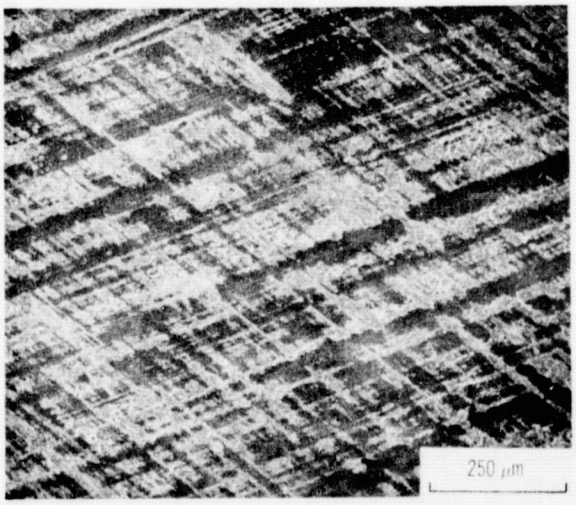


(c) Half specimen/half specimen geometry. Diffusion welded at 1200 K and 1.50MPa for 6500 sec to 0.019 strain. Ground faying surfaces.



(d) Half specimen/interlayer/half specimen geometry. Interlayer from specimen which was diffusion on welded at 1200 K and 1.42MPa for 2150 seconds. Ground faying surfaces.

Figure 8. - Continued.

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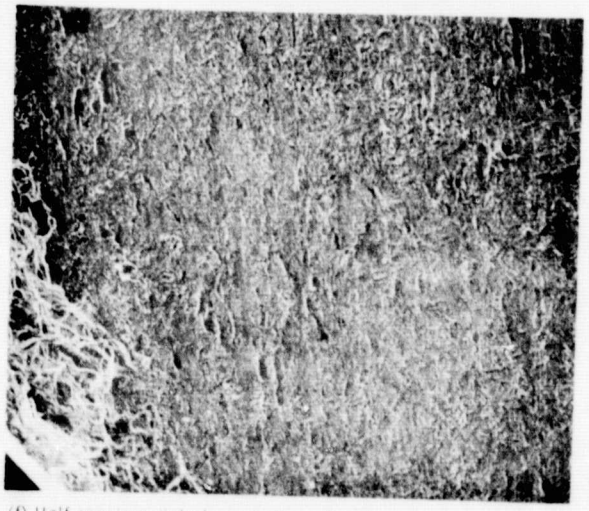
(e) Half specimen/interlayer/half specimen geometry. Interlayer from specimen which was diffusion welded at 1200 K and 1.42MPa for 5900 seconds. Ground faying surfaces.



(f) Half specimen/interlayer/half specimen geometry. Interlayer from specimen which was diffusion welded at 1200 K and 1.42MPa for 12 500 seconds. Ground faying surfaces.

Figure 8. - Concluded.

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